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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

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While a large number of models have attempted to correlate toughness in the ductile/brittle transition regime to events occurring ahead of the crack tip, surprisingly little attention has been paid to events occurring behind the crack front. Fractographic evidence as well as metallographic sectioning of arrested cracks show that the mechanism of rapid crack propagation by cleavage is affected strongly by partial crack-plane deflection which leaves unbroken ligaments in its wake. The tearing of these ligaments by dimple-rupture is the dominant energy-absorbing mechanism. Etch-pit experiments using an Fe-Si alloy show that the crack-tip stress intensity based on plastic zone size is extremely low. It is suggested that the mechanism of crack arrest should be modeled using a sharp crack which is restrained by a distribution of discrete pinching forces along its faces. The same model is applied to crack initiation.

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MICROMECHANISMS AND TOUGHNESS FOR CLEAVAGE FRACTURE OF STEEL

A. R. Rosenfield and B. S. Majumdar
Battelle Memorial Institute
Columbus, Ohio, 43201, USA

ABSTRACT

>A complete understanding of the fracture mechanisms of steel in the ductile/brittle transition region requires analysis not only of crack initiation, but also of crack propagation. This paper reviews micrographic and fractographic experiments that give insight into both phenomena, and suggests a frame-work through which both may be related.

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BACKGROUND

Review of crack initiation

This paper reports the preliminary development of a model of cleavage crack initiation which has its genesis in two Battelle papers which appeared fifteen years ago. In the first paper [1] an analysis of the temperature and loading-rate dependence of K_{Ic} was presented and it was found empirically that:

$$K_{Ic} Y^2 = gS^3 \tag{1}$$

where Y is the yield strength, S is the cleavage strength and g is a proportionality constant.

In discussing Eq. (1), it was suggested that loading of a cracked specimen was accompanied by formation of a plastic zone which generates stress-induced microcracks. This suggestion is consistent with the subsequent acoustic emission results of Khan, et. al. [2], who detected several hundred signals during loading prior to cleavage. Ref. [1] further suggested that cleavage-crack extension occurs when a critical array of microcracks is formed, the variability in K_{Ic} being associated with the stochastic nature of microcracking. The statistical approach to analyzing cleavage data has been developed further by Beremin [3], who has derived a slightly different relation than Eq. (1). In addition, Wallin [4] has shown that the thickness variation of K_{Ic} can be deduced from Weibull (weakest link) statistics.

Shortly after the publication of the first paper [1] , Ritchie, et. al. [5] suggested that cleavage occurs when some critical stress is reached at a microstructurally-significant distance ahead of the crack tip. While initially the significant distance was assumed to be some multiple of grain size [5], it was later suggested [6] that the significant distance corresponded to a most probable distance ahead of the crack tip for carbide triggering of cleavage. Although this theory has been widely accepted, no fractographic observations in support of the model were reported until recently. This is surprising since the cleavage origin can be readily located by tracing tear ridges as illustrated by Figures 1 and 2.
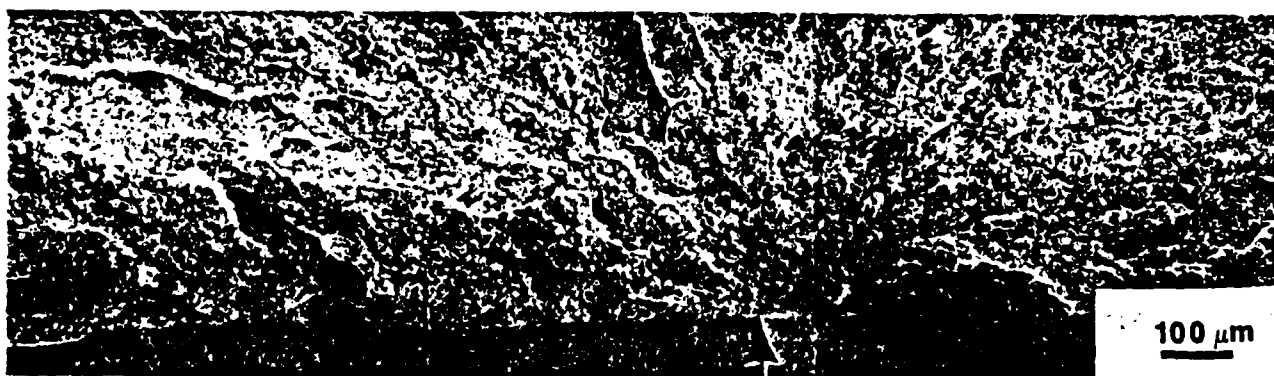

Figure 1. Fracture surface of an A508 steel compact specimen tested in the ductile/brittle transition region (Ref.[9])
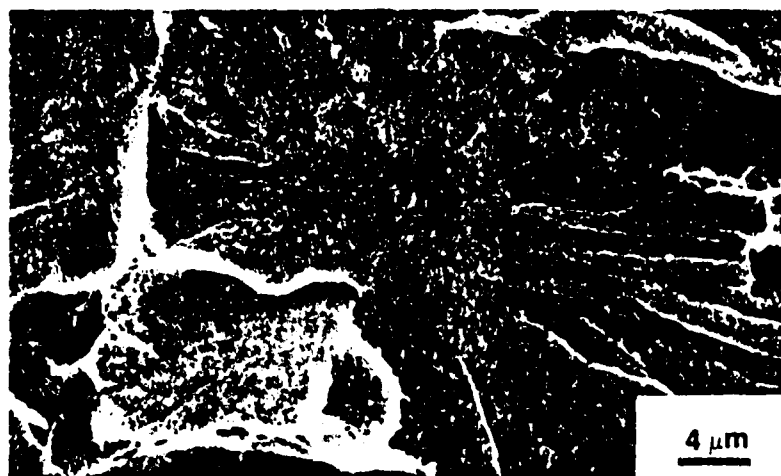

Figure 2. Sulfide particle at cleavage fracture origin in A508 steel. (Ref.[9])

TABLE 1. Selected cleavage-fracture origins

Steel	Origin	Method	Reference
Thoriated Iron	Dislocation Coalescence	Theory	[7]
A508	Dislocation Coalescence	Theory	[8]
Mild Steel	Grain-Boundary Carbides	Theory	[5]
A508	Grain-Boundary Carbides	Fractography	[9]
A533	Prior-Austenite Grain Boundaries (a)	Fractography	[13]
A508	MnS	Fractography	[9]
Weld Metal	Inclusions	Fractography	[10]
Feritic Stainless	Borides	Fractography	[11]
Dual Phase	Martensite Islands	Fractography	[12]

(a) Authors suggest that actual origins are auto-tempered carbides.

Table 1 provides a selection of origins that have either been observed or suggested theoretically [5,7,8,9,10,11,12,13]. Note that while grain-boundary carbides have received the most theoretical attention, the majority of experimental observations have shown a variety of other constituents to be critical to the triggering of cleavage. One possibility for the scarcity of evidence of carbide-induced cleavage is that the carbides in many steels may not be large enough to trigger cleavage [14]. In this respect, our results [9] are of interest since triggering by carbides predominated below the impact-ductile/brittle transition temperature (DBTT) where the maximum tensile stress is relatively high. Sulfide-induced cleavage predominated above the DBTT since the maximum tensile stress was probably not sufficiently large to trigger cleavage.

A later paper [15] contained a detailed discussion of the source of scatter in K_{IC} for cleavage--a marked variability in the distance between the fatigue precrack and the trigger particle. The picture that emerged involves crack blunting accompanied by an expansion of the plastic zone. As the zone expands, it causes local particle fracture. Provided the particle is large enough and the maximum tensile stress (S in Eq.(1)) is high enough, such cracking can result in cleavage failure. If these conditions are not reached before the onset of dimpled rupture, macroscopic ductile fracture is initiated. However, a plastic zone sweeps through the material preceding the growing ductile crack. The stresses and constraints associated with this zone increase with the growth of the ductile crack (because of the increase in J) allowing small particles to trigger cleavage.

At instability, one or a few of the microcracks propagate by cleavage in all directions from the trigger point, which is ahead of either the fatigue precrack or the growing ductile crack. Ebrahimi [16] has presented a slightly different mechanism. Her fractographs also suggest that cleavage microcracks form ahead of the precrack, but that these join the precrack by a shear mechanism instead of by cleavage. Whichever viewpoint is correct, there is a period prior to cleavage failure during which the plastic zone contains an array of microcracks, consistent with the acoustic-emission evidence of Khan et. al. [2]

A method of analyzing experimental records to deduce K_{IC} (the "available energy" method) has also been suggested [9,15,17]. The method is based on the idea

that cleavage initiation and propagation involve elastic unloading of either a blunted precrack or a growing ductile crack. It is argued that only the elastic energy stored in the specimen is available to drive a cleavage crack. Thus the plastic-deformation and crack-growth contributions to the experimentally-measured displacements are eliminated, the specimen being treated as if it were linear elastic and containing a crack length equal to the instability value. This method provides good agreement between compact-specimen and large-cylinder thermal-shock K_{IC} values [15]. The cylinders are sufficiently large that complications due to plasticity are minimized in those experiments. Merkle [18] has offered an alternative method of accounting for plasticity, suggesting that the difference between small and large specimen data is due to constraint differences. This explanation seems unlikely in view of Wallin's analysis cited earlier [4].

An additional observation is that reinitiation of an arrested cleavage crack in a compact specimen provides K_{IC} values more representative of the cylinder data than do either a blunted fatigue precrack or a growing ductile crack [19]. The difference between sharp and blunted cracks may be due to a decrease in the crack tip stress field combined with an increase in the highly deformed region associated with the loss of constraint due to decreasing notch acuity. Both factors need to be combined into a Weibull weakest-link analysis to describe the effect of crack-tip blunting on cleavage [20]. There are also differences in plastic zone size associated with stationary and slowly growing cracks [21], which may be related to the differences in K_{IC} between reinitiation of a sharp cleavage crack and cleavage initiation from the tip of a sharp ductile crack. Finally, it is important to mention that the initiation toughness for the cylinder data were obtained from a series of crack initiation and crack arrest events. Therefore the similarity between the reinitiation data for a compact tension specimen and the initiation data for the large cylinder may be entirely due to characteristics of crack propagation and arrest events, rather than initiation events ahead of a stationary fatigue crack. The model in this paper pertains to the arrested-crack morphology.
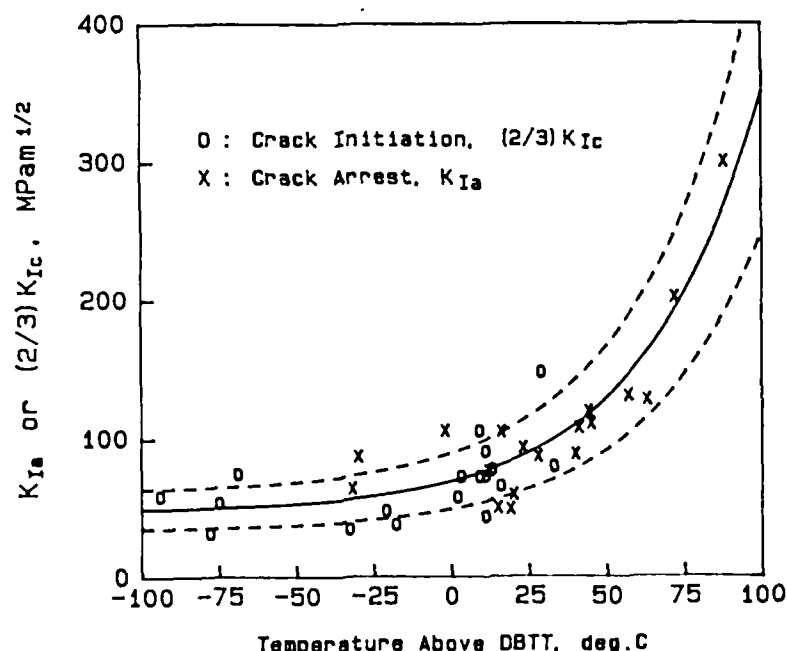

Figure 3. Comparison of crack-initiation and crack-arrest data, thermal-shock experiments in the ductile/brittle transition region.

Review of crack arrest

Developing a microstructurally-based model to explain the experimental observations of crack initiation presents a major difficulty. Some information is available on particle-size distributions [22,23], but not on spacing distributions. The significant departure of the model developed later in this paper from other existing models based on events occurring ahead of the crack tip, is that our model also incorporates microstructural features associated with crack propagation. Knowledge of crack-propagation events are partially obtained from crack arrest studies. Since the stress intensity at crack arrest (K_{Ia}) is close to $(2/3)K_{Ic}$ over a wide temperature range (Fig. 3 [17]) the two quantities may very well be related.

Figure 4 shows that an important feature of propagating and arrested cracks is the occurrence of unbroken ligaments behind the crack tip. In the second of the

<u>Figure 4</u> . Cleavage crack profile of A508 steel showing unbrokenligaments.
· Photograph by P. N. Mincer, Battelle

early Battelle papers [24], ligaments were examined by studying profiles of arrested cleavage cracks in Fe-3% Si. Sections of this steel were etch pitted to reveal local plastic flow. There were three important observations:

(1) The tip of the arrested crack had a very small plastic zone, indicating a very low crack-tip stress intensity after arrest.

(2) Plasticity was concentrated at ligaments behind the crack tip. The ligaments close to the crack tip remained unbroken.

(3) The remainder of the flanks of the crack was apparently dislocation-free, suggesting that the tip of the propagating crack has a very low stress intensity associated with it.

Reference [24] contains an analysis which allows the crack-tip stress intensity to be calculated if the distribution of unbroken ligaments is known. As shown in Figure 5 the model consists of a crack under an applied stress and restrained by pinching forces along its flanks. An interesting sidelight is that the model may well also apply to intergranular fracture of alumina. Crack paths in that ceramic can be quite irregular leading to rubbing of the opposite faces of facets normal to the major crack-propagation direction [25]. This rubbing inhibits crack growth and leads to apparent R-curve behavior (increasing toughness with increasing length of crack travel).
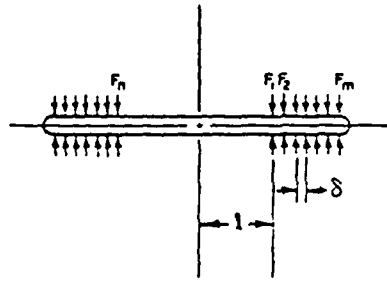
Figure 5.The pinching model of fracture presented in Ref. [24]. Ligaments have been replaced by forces, F_i.

Subsequently [26], arrested cleavage cracks in a bainitic steel (ASTM A508) were examined. Two additional observations were made:

(4) Ligaments fail by a dimpled-rupture mechanism.

(5) The percentage of fracture surface containing dimpled rupture increases as the test temperature increases, accounting for the rise in toughness.

The existence of the combined dimple-plus-cleavage fracture mechanism has been confirmed independently on the same steel [27].

The data in Ref. [24] were analyzed in terms of energy absorption associated with crack propagation. It was found that most of the crack-propagation energy was consumed by ligament formation and rupture. The implication is that the arrested crack tip is well shielded [28], i.e. the actual crack tip stress intensity is significantly less than the stress intensity calculated on the basis of load and crack length.

A SHIELDING MODEL OF CLEAVAGE FRACTURE

Derivation of the model

The present model of cleavage fracture is based on fractographic evidence of uncracked ligaments left in the wake of propagating cleavage cracks. It will be shown that the ligaments can be very effective in shielding the crack-tip stress intensity from the externally applied stresses. The model does not incorporate plasticity influences on crack tip shielding associated with crack-dislocation interactions [28,29], which also contribute to the crack-tip stress intensity.

Consider the Mode I arrested crack of length 2a, as shown in Figure 6(a). It is assumed that the crack has propagated from an initial crack of length much smaller than 2a, and that during propagation the crack has left behind a region of ligaments which bridge the crack surfaces. Let d represent the distance to which the ligaments extend behind each crack tip. Figure 6(b) is a plan view of the crack surfaces, and shows schematically the distribution of ligaments. While fractographic evidence suggests that there is a greater fraction of ligaments immediately behind the crack tip, we shall for the present assume that the ligament fraction is uniform over the distance d. Let f denote the area fraction of ligaments based on their projected area on the fracture surface. There is one other complication: the ligaments are not normal to the crack plane, but rather are inclined at an angle B and seem to fail in a shear mode. This feature is illustrated in Figure 6(c). We shall assume that the ligaments are stretched such that the stresses in them correspond to the yield stress.

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The analysis of the ligament contribution to fracture is fairly straightforward. At a distance x behind the crack tip, the force per unit width exerted by the ligaments on the crack surfaces over an incremental length dx is given by:

$$dF = cfYdx \qquad (2)$$

where c is a factor which accounts for the orientation of the ligaments. Thus in Figure 6(c):
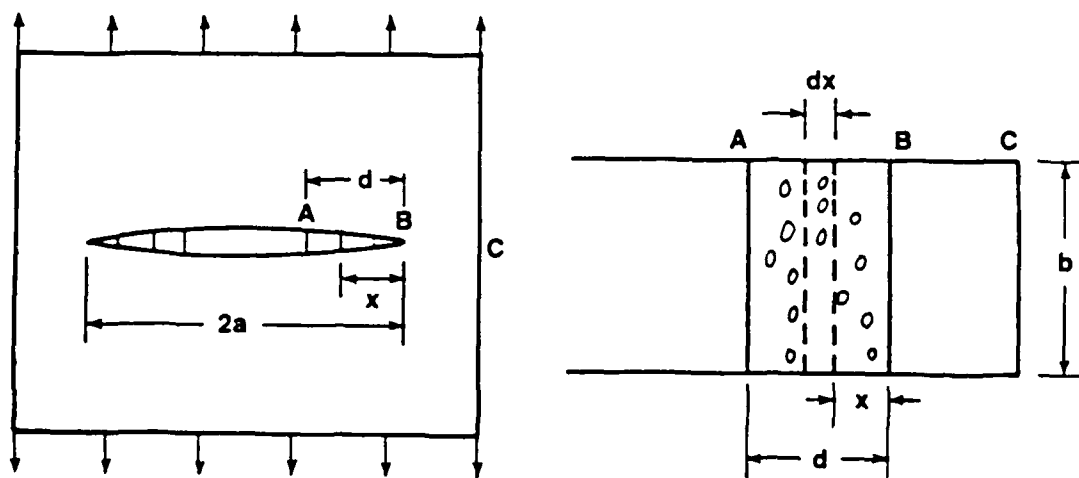
$$c = \tan B /2 \qquad (3)$$

The stress-intensity factor due to this restraining force is given by:

$$dK_L = cfY ((2a-x)/\pi ax)^{1/2} \qquad (4)$$

If d << 2a, integrating between limits 0 and d, the stress intensity contributed by the ligaments is found to be:

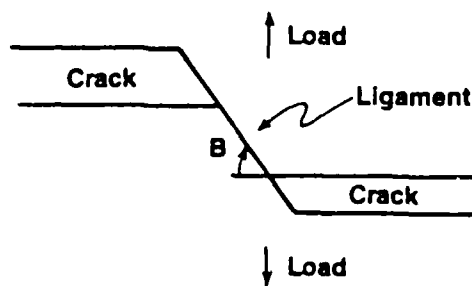$$K_L = (fY \tan B)(2d/\pi)^{1/2}. \qquad (5)$$

Actually, Eq. (5) is a close approximation to the complete result provided d < a.

(a) Schematic of crack

(b) Schematic of fracture surface

(c) Schematic of side view of fracture surface showing ligaments. The crack is propagating perpendicular to plane of paper.

Figure 6. Shielding model for cleavage fracture

Application to crack propagation and arrest

The ligament contribution to the measured stress intensity can be estimated from the data in Table 2. The major problem is estimating tan B since B is close to 90 deg. [24] and its tangent varies rapidly with angle. We have set B=75deg. Evaluating Eq. (5) shows that K_L can account for approximately all of the stress intensity associated with the arrest of a cleavage crack. The size of such other factors as crack branching or plasticity seem to be much less. The result of this analysis is in contrast to most theoretical models (e.g. [30]) which consider dissipative processes occurring only at the crack tip.

Application to crack initiation

Reinitiation of an arrested crack involves two factors: generation of additional ligamentation due to particle cracking and crack-tip plasticity. The ligament contribution to the stress intensity at cleavage-crack initiation is probably more modest than for crack arrest. Initiation may best be understood by considering that the ligaments give rise to an R-curve behavior. Thus, as microcracking occurs ahead of the main crack, more and more ligaments are left behind in the wake, so that higher loads become necessary to advance the crack front. Ebrahimi's [16] micrographs suggest that ligament separation occurs along Prandtl slip lines so that B = 45 deg. Secondly, d can be set equal to the plastic zone size:

$$d = (1/6\pi)(K_{Ic}/Y)^2. \tag{6}$$

TABLE 2. Crack-arrest and ligament data for steels

Steel	Test Temp., deg. C	Ligamented Fraction, f	Length d, mm.	Yield Strength, Y, Mpa	K_L (a) MPam**.5	K_{Ic}, MPam**.5	Ref.
Fe-3%Si	-75	0.15	10	650	29	33	[24]
A-517	-196	0.20	2.5	1230	37	31	[24]
A-508	-7	0.14	103(b)	500	67	72	[26]
A-508	22	0.17	83(b)	500	73	114	[26]
A-508	43	0.22	72(b)	500	88	138	[26]

(a) In all cases the ligaments are assumed to be oriented at 75 deg. to the crack plane.
(b) The ligamented length was assumed equal to the total crack-propagation length.

The value of d given in Eq. (6) probably represents a lower limit on the microcrack zone, since the data on Fe-3% Si [24] shows that the microcracked zone can be as much as three times larger. Partial compensation for short ligament zone is provided by a larger fraction for initiation than arrest; we have assumed that $f = 1$. Combining the above factors, Eq. (5) becomes:

$$K_L = K_{Ic}/5. \tag{7}$$

Comparing this estimate with experiment is difficult. Recalling the result in Figure 3 that $K_{Ia} = 2K_{Ic}/3$, Eq. (7) suggests that the ligament contribution to toughness represents only about half of the difference between the two measured quantities. However, the situation is complicated, since most of the data in Figure 3 was obtained on cracks that propagated, arrested, and reinitiated. Therefore these data reflect two possible ligament contributions: the contribution of ligaments formed during propagation and the contribution of additional ligamentation associated with crack initiation. The data analysis in [19] shows that the available energy energy correction typically reduces the reported K_{Ic} values by 22%. Assuming that this correction adequately accounts for the effect of screening due to plasticity, the remainder of the discrepancy between K_{Ic} and K_{Ia} can be explained. Again there is a complication in that this comparison is based on the existence of a fatigue precrack instead of an arrested cleavage crack. It may finally be pointed out that in tests where crack jumps prior to crack arrest were only of the order of 1-10 mm, the K_{Ia} values and K_{Ic} values were almost equal. This would seem to suggest that ligament contribution is the dominant factor in crack initiation of such specimens.

CONCLUDING REMARKS

The thrust of this paper is that analysis of experimental records to provide K_{Ic} values requires an understanding of the micromechanism of cleavage fracture. While most models attempt to correlate toughness with events occurring ahead of the crack tip, this paper illustrates that events occurring behind the crack tip can play a role. This understanding is particularly important when extrapolation from small-specimen to large-structure data is required. Fortunately, since crack-arrest toughness, K_{Ia}, appears to be the most conservative measure of resistance to cleavage fracture and there is an extensive body of data on this quantity, an engineering solution to specifying "lower-bound" toughness is on hand, as illustrated by the scatter band in Figure 3. However, it must be strongly emphasized that almost every result in this paper (and in much of the recent research), has been performed on low-alloy reactor-pressure-vessel steel and that the quantities used to evaluate the ligament model may not be generally applicable to all steels. In this context, it is useful to note that much of the earlier work was done on mild steel, which contains very large carbides and has a very different microstructure from many steels commonly used today.

ACKNOWLEDGMENTS

Availability Codes

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